Periodical

SUPPLEMENT TO THE ENGINEER, APRIL 27, 1934.

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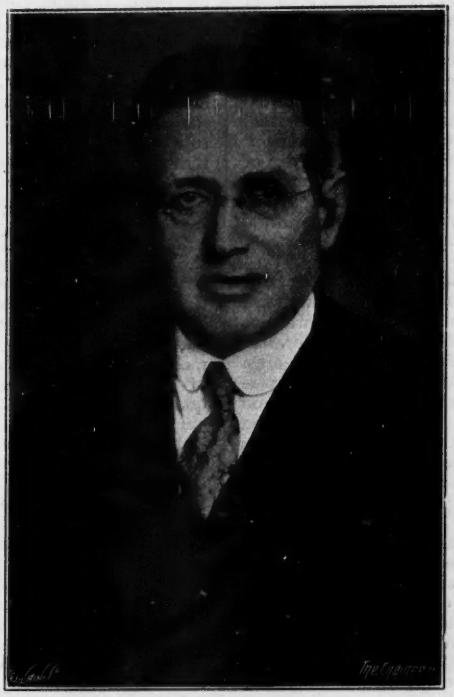
The Metallurgist.

Supplement to The Engineer, April, 1934.

Dr. Walter Rosenhain.

IT is fitting that there should be recorded in THE METALLURGIST the death of Dr. Walter Rosenhain on March 17th, for, as already mentioned in THE ENGINEER of March 23rd, this Supplement was established on a proposal made by him, and was directed by him from its first appearance until his death.

As all metallurgists know, some at the cost of sore heads, Rosenhain might be described not unfittingly as a militant metallurgist. Holding strong views, he fought for them stoutly and was a dangerous opponent in debate. But his battles were not always with his brother metallurgists, and he never showed greater zeal and energy than when he was contending for the development of metallurgical research — and indeed research of all kinds. Practically all the editorials — the two first articles in each issue of THE METALLURGIST — were written by him, and his touch must have often been recognised or suspected by those not in the secret of their authorship. In the early days, before he had learnt by experience that editing calls for restraint, his manuscripts gave some anxiety to the Editor-in-Chief, and for a few months he was not over tolerant of the modifications which had to be insisted upon. But he rapidly acquired the art of saying effectively what he wanted to say without running the risk of antagonising a large proportion of his readers.



DR. WALTER ROSENHAIN

To the work he brought a measure of knowledge and experience which is possessed by few. His duties at the National Physical Laboratory had put him into contact with many phases of metallurgy, and with the help of the equipment provided at the Laboratory he had either personally or through his assistants carried out endless experiments and tests. He was therefore in a position to speak always intelligently and in most cases with real weight on a great diversity of metallurgical questions. Added to this direct knowledge was his familiarity with several foreign languages, which threw open the literature of other nations to him and placed him in a position to converse freely with foreign workers. Hence the comments which he was fond of adding to the excerpts from foreign sources which have always been a feature of THE METALLURGIST, were marked by the true critical spirit.

Few men possess the extraordinary power of exposition which was exhibited by him. It was seen best when he was called upon to address audiences not familiar with the jargon of metallurgy. On such occasions he would handle a long and detailed report, and in the course of half-an-hour or so lay all the essentials so clearly before his hearers that everyone had the comfortable feeling that he—the hearer—really understood the subject! When discoursing to his fellow metallurgists this gift (being not so strongly called for) was less patent, but he never failed to be an interesting, an instructive, and an arresting speaker.

Such men as he are rare indeed, and it will be long before his place is completely filled again. He brought his talents to bear just at the time when they were needed. The importance of the science of metals had to be impressed upon the users, of whom the greater number are engineers. To that task he devoted his energies, and it was in the pursuit of it that he enlisted the co-operation of THE ENGINEER in the publication of a journal which would go direct to engineers and be read by all metallurgists in iron and steel and non-ferrous factories and engineering works. Hence THE METALLURGIST; a work which must be numbered with all the other things that he did to foster, encourage, and promote the progress of the science to which he devoted his life.

bearing alloys, and while adherence to the limits imposed by British and American specifications has resulted in little trouble being experienced with automotive engines, it is anticipated that troubles might be experienced with the present trend towards increased horse-power and higher working temperatures, and it has even been predicted that tin-base alloys for automotive engines will eventually be abandoned. The effect of lead at various temperatures in amounts up to 14 per cent. was investigated on three alloys, the compositions of which are indicated in Table II. The tests were made on small cast

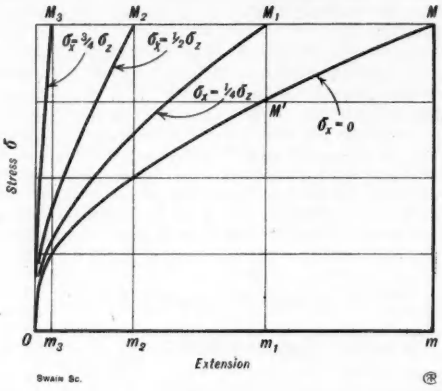
TABLE II.

	Alloy No. 1.	Alloy No. 2.	Alloy No. 3.
Tin, per cent.	80·0	89·0	91·0
Antimony, per cent.	15·0	7·5	4·5
Copper, per cent.	5·0	3·5	4·5
Lead, per cent.	0 to 14 by addition		

specimens which during testing were in direct metallic contact with a heavy mass of steel in a bath of low conductivity oil, the heat being supplied from immersed heating coils. Serious temperature fluctuations were avoided, and it was found possible to test the whole series of specimens under practically identical conditions. Four specimens were cast for each loading condition and one hardness impression was made on each for a given temperature increment. The temperature was maintained approximately constant for one hour before the tests were made. With alloys Nos. 1 and 2 the addition of 4 per cent. lead and with alloy No. 3 the addition of 3 per cent. lead, gave the greatest increase in hardness for all temperatures up to 100 deg. Cent. The addition of 8 per cent. of lead showed an increase in hardness of 2 points at room temperature, but none at 100 deg. Cent. Up to temperatures of 183 deg. Cent., the addition of 3, 4, and 8 per cent. lead does not result in any important falling off in hardness, but the probable hardness values at higher temperatures have been obtained by extrapolation of the straight line relationships derived from plotting on a logarithmic scale hardness values for temperatures up to 100 deg. Cent. These values and the actually determined values on the three alloys without lead additions and containing lead in amounts which give rise to the greatest hardening effects, are given in Table III. It was observed that increasing accuracy in the hardness determinations was obtained if the load was such as to avoid giving rise to "cratering" of the Brinell impression. Hardness values obtained with loads of 250 kilos. and 500 kilos. did not differ by more than about 1 per cent. when the diameter of the impression was of the order of 5 mm., but with a 6 mm. impression showing a tendency to crater the differences were of the order of 6 per cent.

Effect of Multi-Axial Stressing on the Ductility of Metallic Materials.

THE most commonly accepted criteria of the ductility of a metal are the elongation and reduction of area, as determined by the tensile test. There is, however, conclusive evidence that these criteria are not of general application, irrespective of the condition of stress, but are valid only for the conditions under which they are measured. The elongation of a tough metal, for example, is determined by the particular degree of distortion at which contraction sets in and the maximum loading is reached. The reduction of area, on the other hand, is characteristic of the state of distortion at which rupture begins in the smallest section of the contracting test piece as a result of the multi-axial condition of stress there existing.



EFFECT OF MULTI-AXIAL STRESSING

Ludwik has shown theoretically how three-dimensional stressing reacts unfavourably on ductility. Thus if the curve O M above represents the stress-extension curve of a particular material under a purely axial stress, f_z, the effect of perpendicular stresses f_x of amount $\frac{1}{4} f_z$, $\frac{1}{2} f_z$, and $\frac{3}{4} f_z$ is to change the curve O M to O M_1, O M_2 and O M_3 respectively. Considering the curves O M and O M_1, the axial stress m M′ is necessary to produce an extension O m_1 when $f_x = 0$, whereas when $f_x = \frac{1}{4} f_z$ the axial stress necessary to produce the same extension is increased to $m_1 M_1$. On the other hand, it is the difference of

TABLE III.

Lead, per cent.	Method of securing Brinell hardness.	Alloy No. 1.			Alloy No. 2.			Alloy No. 3.	
		151° C.	179° C.	193° C.	151° C.	178° C.	193° C.	125° C.	175° C.
0	Extrapolated	8·0	5·9	5·2	6·2	4·8	4·5	6·9	3·8
	Determined	8·4	5·6	5·0	6·4	4·3	3·7	8·0	4·4
3	Extrapolated	—	—	—	—	—	—	7·5	4·1
	Determined	—	—	—	—	—	—	8·6	4·4
4	Extrapolated	8·7	6·4	—	6·9	5·1	—	—	—
	Determined	7·5	4·3	*	6·6	3·8	*	—	—

* At this temperature the specimens containing lead crumbled under the applied load.

principal stresses $f_z - f_x$ that is the true measure of the ductility or extension, and this is always the same for any given extension as the axial stress f_z when $f_x = 0$. Alternatively, regarding the matter from the point of view of a constant stress to produce failure, which for the present purpose we may consider is given by the upper horizontal line of the diagram, we see that the greater the transverse stress, the less the extension at which failure occurs. Hence under three-dimensional loading, even materials characterised by extreme ductility in the tensile test may exhibit brittleness.

It appears also that the ductility may be considerably diminished in cases of multi-axial loading, even when the transverse stress along a principal axis is zero or negative, and in *Z.V.D.I.* for December 23rd, 1933, Messrs. E. Siebel and A. Maier describe experiments to furnish information on such cases. Their tests comprised measurements of the changes of dimensions of tubular test pieces of different steels subjected to axial tension or compression, and, simultaneously, to tangential stress f_t by the introduction of oil under high pressure to their interiors. The range of stresses varied from pure tension (f_z finite and $f_t = 0$) through a range of two-dimensional stresses to pure transverse tension ($f_z = 0$, f_t finite). The radial stress f_r was always adjusted to be zero. Tubes of steel St 37 and of steel St 60, both annealed and heat treated, and of brass Ms63 were used in the investigation. In every case they were stressed to failure for each condition of loading, and measurements were made of the maximum axial and circumferential extensions (e_z and e_t respectively). It should be added that a given stress condition was defined by the ratio of the stresses f_z and f_t, so that comparison was possible between the results of the different materials, although the absolute values of the stresses in any given condition were different in different materials.

It was found that the axial extension e_z decreased continuously from the case of pure axial tension to a small negative value for pure transverse tension, while the tangential extension e_t went through a continuous change in the reverse direction, being negative for purely axial loading and positive for the transverse condition. When the loading was such that f_z was equal to f_t, then the two extensions e_z and e_t were also of equal value.

If the value of the greatest positive extension at fracture (either e_z or e_t) found for any loading condition is taken as a measure of the ductility under that condition, the results show that the ductility of all the materials was greatest for purely axial stress, dropped to a minimum when axial and transverse stresses were equal, and rose again continuously to the condition of pure transverse tension.

For St 37 the maximum value of e_z was 74 per cent., whereas the ordinary tensile test on a bar gave 93 per cent. The minimum value when f_z was equal to f_t was 12 to 15 per cent. Similar variations were found for the other materials ; in no case was the maximum value of e_z as great as that found in the tensile test with a flat specimen. It is noteworthy also that in the case of the heat-treated steel St 60, the ductility for the case $f_z = f_t$ was practically zero.

Interesting differences in the types of fracture under different stressing conditions were observed. As long as the axial stress exceeded the transverse stress, the plane of fracture was approximately perpendicular to the axis of the tube. Similarly, the fracture took the form of a longitudinal tear in the walls of the tube when the transverse stress was greater than the axial. When the two stresses were equal, fractures in both directions were observed. The plastic materials showed pronounced contraction of area only under uni-directional tension along the axes of the specimens. In the case when the circumferential stress was half the axial, the tubes invariably remained cylindrical right up to fracture without showing any appreciable sign of radial contraction. It is particularly interesting to observe that the fractures had a distinctly brittle appearance when they occurred longitudinally under conditions in which the circumferential stress was appreciably greater than the axial, and that there was then no sign at all of any radial contraction. In a few cases of this type of loading, the tube was more or less barrel-shaped at failure.

After a theoretical discussion of the data obtained, the authors give a brief account of some tensile tests carried out on solid cylindrical specimens subjected to high external pressure—some hundreds or even thousands of atmospheres. The transverse stress in this case was evidently of opposite sign to that in the tubular specimens under internal pressure, and the tests were made to see if these conditions led to the reverse effect, *i.e.*, an increase of apparent ductility, to that previously found. The experiments showed that this was indeed the case, and also that the breaking stress was increased by the pressure. The uniform extension was practically independent of the external pressure, as was to be expected, but the contraction at fracture, and hence the local extension increased with the pressure. The latter result was no doubt due to the fact that the transverse stresses that are normally set up in the plastic cone (*Fliesskegel*) during the tensile test were reduced or even reversed in sign by the external pressure.

The fact that two or three-dimensional stresses reduce the ductility to a fraction of its value in the normal tensile test is obviously one that cannot be disregarded in actual practice. Tubular constructional members subjected to internal pressure, for example, should only be made of materials that exhibit in the tensile test a degree of ductility sufficient to allow for the possible reduction. Otherwise, there is a danger that temporary overloads may cause fracture without enabling the excess stress to be taken up, as it might be with a more ductile material, by a certain amount of permanent distortion. The ideal, of course, would be to test the members in question under internal pressure *in situ*.

Columbium in Chromium-Nickel Steels.

INTER-GRANULAR corrosion in chromium-nickel austenitic steels is a well-known type of defect, which in recent years has given rise to a good deal of research, whilst much theorising has been indulged in regarding its cause. Many methods have been suggested for overcoming the trouble, these generally involving the addition of other elements. At present titanium is most commonly used as the extra alloying addition. An investigation has been carried out by F. M. Becker and R. Franks and reported in Technical Publication No. 515 of the American Institute of Mining and Metallurgical Engineers on the resistance to corrosion of plain chromium-nickel steels and similar steels containing additions of columbium, tantalum, and other metals, after being maintained for long periods at temperatures ranging from 300 deg.

to 900 deg. Cent. The corrosion tests were carried out in a boiling solution containing 6 per cent. copper sulphate and 10 per cent. sulphuric acid, while for some tests a boiling solution of 65 per cent. nitric acid was used. All the steels tested were cast into 2in. or 8in. square ingots and rolled into 1in. round bars or sheet of various thicknesses. Inter-granular tests were carried out on samples of 20-gauge sheet quenched from 1150 deg. Cent. The samples were heated at 300 deg., 400 deg., 475 deg., 550 deg., 650 deg., 750 deg. and 850 deg. to 900 deg. Cent. for one hour, twelve hours, three days, five days, two weeks, one month, two months, and in some instances for five months. They were then boiled in the acid copper sulphate solution for at least 200 hours, unless failure occurred earlier. The tensile and impact specimens were machined to standard dimensions from 1in. round bars and either quenched or air cooled from 1150 deg. Cent. prior to testing. The results obtained with straight chromium-nickel steels are indicated in Table No. I., from which it will be noted that all the steels containing between 12 and 27 per cent. chromium, 6 to 25 per cent. nickel and relatively low carbon content could be disintegrated after holding at temperatures between about 400–750 deg. Cent., although the steels having the higher chromium and nickel contents with relatively lower carbon were subject to disintegration between a narrower range of temperature. The susceptibility of these steels to inter-granular corrosion was not found to be accompanied by a decrease in toughness. Holding at higher temperatures resulted in a loss of toughness, but the steels so treated were not subject to intercrystalline corrosion. There are thus two problems arising in the use of austenitic chromium-nickel steels at elevated temperatures, although at 650 deg. Cent. both defects occur together and may intensify each other.

As will be seen from Table No. II., the addition of columbium to the steels yielded very satisfactory results, the presence of approximately four times as much columbium as carbon diminishing the range of temperature through which the steels were attacked at the grain boundaries. Increasing the columbium to carbon ratio to about 7 : 1 effected a marked improvement, although the steel could still be disintegrated, but when there was at least ten times as much of columbium as carbon present the steels could not be disintegrated after exposure for either long or short periods at all temperatures between 300 deg. and 850 deg. Cent. Except in the case of steels of very low carbon content, the presence of columbium does not counteract the loss of toughness after exposure within the range 650–900 deg. Cent. No thermal changes were detected in columbium steels such as were observed by the special experimental technique which revealed their presence in corresponding straight chromium-nickel steels. Typical physical properties of chromium-nickel steels containing columbium are given in Table No. III. These steels were found to possess excellent resistance to salt spray in tests of 750 hours' duration and equally

TABLE I.—*Tests on Plain Chromium-nickel Steels of Varying Compositions (All Samples Quenched from 1150 deg. Cent. before Testing).*

Cr,* per cent.	Ni, per cent.	C, per cent.	Approximate temtures within range 300°–850° C. that cause subsequent disintegration in boiling acidified copper sulphate, deg. cent.	Approximate temtures within range 300°–850° C. that cause decrease in Izod impact values, deg. cent.
12·74	6·27	0·03	400 to 550	Not tested
12·82	12·80	0·04	400 to 550	Not tested
12·52	25·44	0·05	400 to 550	Not tested
18·23	8·49	0·04	400 to 550	Not tested
17·47	9·27	0·07	400 to 650	650† to 850
18·45	8·56	0·12	400 to 750	550† to 850
18·40	9·37	0·19	400 to 750	550 to 850
18·37	14·12	0·09	400 to 550	650† to 850
20·42	11·87	0·09	400 to 550	650† to 850
22·25	15·06	0·04	475 to 550	Not tested
22·32	20·38	0·05	475 to 550	750 to 850
27·32	25·19	0·07	475 to 550	Not tested

* These steels also contained normal amounts of manganese, silicon, phosphorus, and sulphur.

† Holding for the same period at temperatures near the lower limit of this range causes only a slight decrease in toughness.

TABLE II.—*Corrosion Tests on Chromium-nickel Steels Containing Columbium (All Samples Quenched from 1150 deg. Cent. before Testing).*

Cr,* per cent.	Ni, per cent.	Cb, per cent.	C, per cent.	Approximate temperatures within range 300°–850° C. that cause subsequent disintegration in boiling acidified copper sulphate, deg. Cent.
18·39	8·39	0·37	0·14	475 to 650
18·49	8·56	0·44	0·10	550 to 650
18·28	9·42	0·51	0·07	550
18·36	9·38	0·79	0·13	650
18·33	9·16	1·51	0·18	550
18·75	9·02	1·00	0·11	650
17·81	9·18	0·75	0·07	None found
18·63	9·27	1·25	0·12	None found
18·75	9·14	2·13	0·12	None found
22·87	12·43	1·50	0·14	None found
26·06	20·44	1·51	0·14	None found

* These steels also contained normal amounts of manganese, silicon, phosphorus and sulphur.

TABLE III.—*Physical Tests on Chromium-Nickel Steels Containing Columbium.*

Cr,* per cent.	Ni, per cent.	Cb, per cent.	C, per cent.	Condition of metal.	Yield point, lb. per sq. inch.	Max. stress, lb. per sq. inch.	Elongation, per cent. in 2in.	Reduction of area, per cent.	Izod impact, foot-lb.	Brinell hardness.†
18·28	9·42	0·51	0·07	Quenched from 1150° C.	36,000	89,500	60	72	120	137
17·81	9·18	0·75	0·07	Quenched from 1150° C.	35,000	91,700	59	67	120	143
18·63	9·27	1·25	0·12	Quenched from 1150° C.	35,000	91,700	58	66	103	143
18·63	9·27	1·25	0·12	Air cooled from 1150° C.	41,000	92,750	57	64	102	146
18·52	9·00	2·13	0·15	Quenched from 1150° C.	41,000	91,200	51	64	94	163
18·52	9·00	2·13	0·15	Air cooled from 1150° C.	36,000	89,500	51	63	98	153
22·87	12·43	1·50	0·14	Quenched from 1150° C.	55,500	91,000	42	58	97	146
22·87	12·43	1·50	0·14	Air cooled from 1150° C.	52,000	91,000	47	64	98	149
25·06	20·44	1·51	0·14	Quenched from 1150° C.	48,000	90,500	44	63	86	146

* These steels also contained normal amounts of manganese, silicon, phosphorus, and sulphur. † 3000-kilos. weight.

satisfactory results were obtained when boiling in 65 per cent. nitric acid. When columbium is present in the correct ratio to the carbon content, chromium-nickel steels can be satisfactorily welded by either the oxy-acetylene or arc methods. Protection is obtained on material not thicker than $\frac{3}{16}$ in. from localised failure in the weld or adjacent zone when the carbon does not exceed 0·10 per cent. and when the columbium is present to the extent of four or five times the carbon content. The required columbium to carbon ratio is raised to 6 or 7 : 1 when the carbon content reaches 0·13 per cent. The fact that little columbium is lost in welding favourably affects the results obtained, and little difference is observable between resistance to corrosion of arc or oxy-acetylene welds with the good welding technique which is required. When resistance of welds to corrosion at elevated temperatures is needed both the filler rod and the material being welded should contain at least ten times as much columbium as carbon.

In making the steels, ferro-columbium of low carbon content and containing from 50 to 60 per cent. columbium should be added through the slag to the thoroughly deoxidised steel just before pouring, after allowing time for the columbium to alloy. With this procedure recoveries of 80 to 90 per cent. of the columbium in the finished steel can be realised. Columbium steels have been fabricated under normal conditions into bars, rods, plates, sheets, strip, and seamless tubes. The initial hot working temperatures should not be much in excess of 1200 deg. Cent. and the annealing and pickling processes for the plain 18 per cent. chromium, 8 per cent. nickel steel are suitable. When columbium steels after quenching are held at temperatures up to 650 deg. Cent. for normal periods, no evidence is obtained at magnifications of 2000 of any precipitation at the grain boundaries, which occurs in the plain 18 per cent. chromium 8 per cent. nickel steel when similarly treated, and it is to this difference that the effectiveness of columbium in inhibiting grain boundary attack is ascribed.

Zinc-Base Pressure Die Castings.

THE progress of mass production has been materially assisted by the pressure die-casting process. By means of the modern die casting machine it is possible to produce, at very low cost, castings of great complexity to very fine limits of tolerance. By the use of such castings, which often require no subsequent machining, the cost of the article is reduced to a remarkable extent. On the other hand, the moulds necessary for the production of these castings are very costly, and generally, in order to recover this cost, the output of each mould must be large. The production and design of the mould is a highly skilled art, and the designer of the article to be produced must be thoroughly conversant with the possibilities and limitations of the process, if really satisfactory results are to be obtained.

The selection of the alloy employed for the process is extremely important. An interesting paper on this subject appears in the December number of *Metal Progress*, summarising some valuable work on the zinc-base alloys recently carried out under the auspices of the American Society for Testing Materials. The zinc-base alloys are cheap and strong, and cast extremely well, but the one most favoured at the time the investigation was started was found to fail under the very exacting corrosion test employed, that of exposure to water vapour at about 90 deg Cent. This alloy contained 3 per cent. copper and 4 per cent. aluminium, the remainder being zinc containing not more than 0·25 per cent. impurities. This alloy when submitted to the steam test failed by intercrystalline corrosion. Considerable improvement was observed as the grade of zinc employed was improved, and it is now universal practice to employ zinc of the very highest grade, 99·99 per cent. pure. This emphasises the important effect that very small amounts of impurity may exert on the quality of material.

The next important step forward was made when it was found that the addition of magnesium in quantity little more than a trace prevented this intercrystalline corrosion to a remarkable extent. The resulting alloy was known as Alloy XXI. and the following tentative standard specification was issued for the pressure cast test piece made from it :—

Composition.	Alloy XXI.	Alloy XXIII.
Copper	2·5–3·5 per cent.	0·1 per cent. max.
Aluminium	3·5–4·5 per cent.	3·5–4·5
Magnesium	0·02–0·1	0·03–0·08
Iron	0·1 max.	0·1 max.
Lead	0·007 max.	0·007 max.
Cadmium	0·005 max.	0·005 max.
Tin	0·005 max.	0·005 max.
Zinc	Remainder	Remainder
Properties as Cast ; Minimum Average of Five Test Pieces.		
Tensile strength, tons per square inch	20	16
Elongation on 2in.	2·0 per cent.	3·0 per cent.
Charpy impact, ft.-lb.	6	12
Properties after Ten Days in Water Vapour at 200 deg. Fah. (about 90 deg. Cent.).		
Tensile strength	13 min.	13
Elongation on 2in.	0·5 per cent. min.	1·5 per cent. min.
Charpy impact	0·75 min.	12·0 min.
Maximum dimensional change	0·0025in. per in.	0·001in. per in.

It will be seen from this table that while the tensile strength of the alloy is not very seriously lowered by the steam test, the ductility as measured by elongation and the resistance to impact are very seriously reduced. In the present state of knowledge it is not possible to forecast with any exactness the behaviour of the material when in normal use from the results of the accelerated test. It is realised, however, that the test is very severe, and represents the equivalent of years of normal service. Some long-period tests were made to gain information on this point. It was found that, as a result of a year's exposure at different stations, the reduction of resistance to impact was generally less than 50 per cent. However, in the humid tropical atmosphere of Key West and the corrosive atmosphere of New Kensington, Pa., the reduction was greater. Alloy XXI. has been extensively used commercially and its practical value has been established. Examples are given of its use as a material for gears and pinions, &c., and as mounting plates for electrical clocks where permanence of dimensions is a property of extreme importance.

Good as Alloy XXI. has proved to be, a further improvement has been made. Earlier experiments made on alloys containing aluminium, but no copper,

had shown that they were too unstable to be of practical value. It was found later, however, that if 99·99 per cent. zinc was employed, the alloy containing 4 per cent. aluminium and not less than 0·03 per cent. magnesium was tougher and also more resistant to the steam test than Alloy XXI. The alloy developed in this way is known as Alloy XXIII. The tentative specification covering it is given in the table above. It will be seen from the table that the mechanical properties of the material are remarkably little altered by the steam test. The maximum dimensional change specified is particularly remarkable. The reason for this very low figure appears to be that the alloy first contracts and then very slowly expands under the test. It is suggested that the use of Alloy XXIII. should greatly enlarge the field for zinc-base pressure die castings, which may now be favourably compared with cast brass or malleable iron in toughness and general permanence.

Gongs.

No. II. (*continued from March 2nd issue*).

EXPERIENCE in manufacture had indicated the value of tempering the finished gong at a low temperature, and a series of experiments was carried out in which gongs made from 0·63 per cent. carbon steel were tempered for various times and at different temperatures. The results are illustrated in Fig. 2, from which it is clear that 300 deg. Cent. is the best tempering temperature as regards the effect on the duration of the note, and also that no advantage results from continuing the tempering for longer than fifteen minutes. Table II. (*ante*) shows how the yield point and tensile strength of the cold-worked material are raised by tempering.

The explanation of these effects of tempering cannot

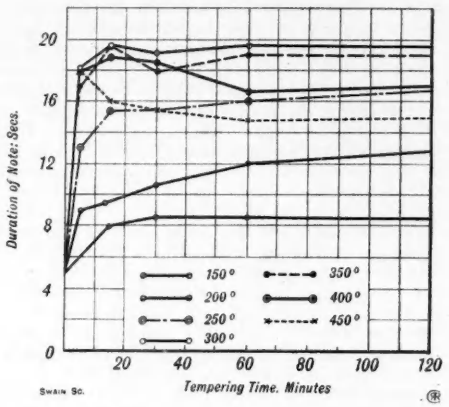
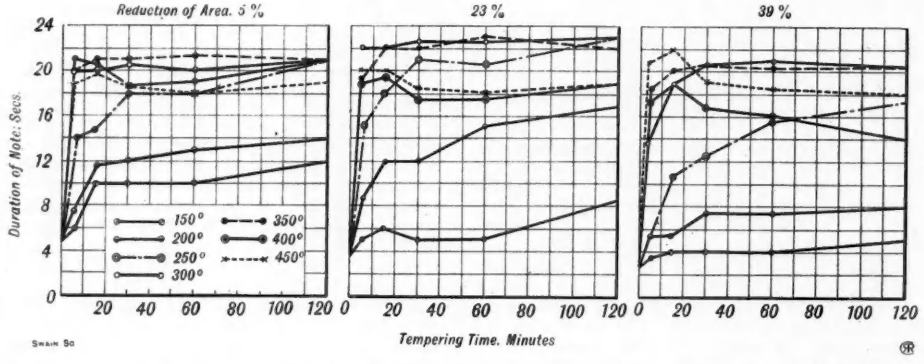


FIG. 2—*Effect of Tempering on Duration of Note: Gongs made from Cold-drawn 0·63 per Cent. Carbon Steel.*

present instance. Support for the view that internal stresses are released by the tempering procedure employed in treating gongs is obtained from X-ray reflection spectrograms of the wires, in which a marked sharpening of the lines is evident after tempering.

Figs. 3, 4, and 5 summarise the changes in note duration with different tempering temperatures in a 0·63 per cent. carbon steel which had been cold drawn by varying amounts, the original diameter being 5 mm. With tempering temperatures of 300 deg. and 350 deg. Cent. variation of the degree of cold work from 5 to 39 per cent. reduction in area had little effect, but with lower temperatures the adverse effect of heavier reductions was not so readily eliminated as was that of light reductions. This again is



FIGS. 3, 4, AND 5—*Effect of Degree of Cold-work and Tempering on Duration of Note: Gongs made from 0·63 Carbon Steel.*

easily be associated with a precipitation process, and, moreover, they are not peculiar to ferrous materials; drawn wires of copper and nickel show the same phenomena. It is more probable that the cause is to be found in the release of internal stresses, but in this event the stresses concerned must be limited to those which cause only a slight distortion of the space-lattice, since the release of "macroscopic" internal stresses requires considerably higher temperatures and longer times than were found to be effective in the consistent with the view that the improvement in note duration on tempering is the result of relieving internal stresses.

The influence of carbon content was studied in a series of five gongs made from wires containing from 0·01 to 1·1 per cent. carbon. In each case a rolled wire of 5 mm. diameter was drawn to 4·6 mm., annealed at 680 deg. Cent., and drawn to 3·6 mm. diameter. All the wires showed a microstructure consisting of spheroidised cementite in ferrite. The

effect of tempering is seen in Fig. 6. In general, increase in the carbon content gave a shorter note in the untempered gongs, and a longer note in those which had been tempered at 300 deg. Cent., although increase in carbon content beyond 0·63 per cent. had no further effect, except in producing a fuller note.

Since the duration of the note from a gong must be

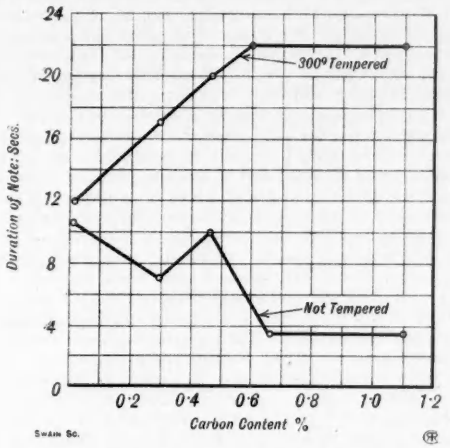


FIG. 6—*Effect of Carbon Content on Duration of Note in Tempered and Untempered Gongs made from Cold-drawn Wire.*

dependent ultimately on the damping capacity of the steel, it was desirable to obtain data on this property by direct mechanical means, to provide a comparison with the method of determining the duration of note. Unfortunately all the available methods of ascertaining the damping capacity imposed stresses approaching the yield point of the material, which

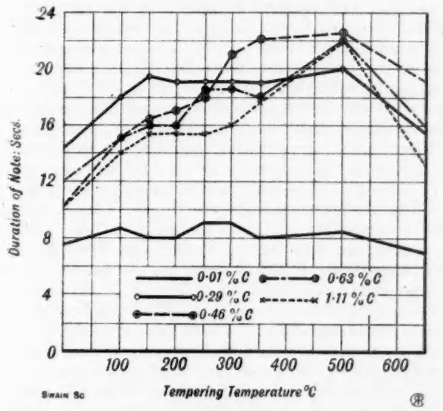


FIG. 7—*Effect of Tempering on Duration of Note in Gongs made from Oil-hardened Wire.*

are, of course, of a totally different order of magnitude from the small stresses imposed in sounding a gong. With this limitation in mind, however, determinations were made on an Esau-Voigt alternating stress machine, the dimensions of the test piece being selected so that its characteristic vibration frequency was about 450 per second, *i.e.*, within the range of the audible notes of a gong. The specimen was set in vibration at its characteristic frequency by electro-magnetic excitation, and by suitable control, after loading to 40 kilos. per square millimetre, the logarithmic decrement was determined. Three specimens were used in these tests: one made from 1·1 per cent. carbon steel annealed at 680 deg. Cent. before cold drawing from 16·5 mm. diameter to 15 mm., another of the same material annealed at 760 deg. Cent. before cold drawing to the same extent, and a third, of 0·45 per cent. carbon steel, similarly cold worked after annealing at 680 deg. Cent. The microstructure of the first specimen was typically spheroidised cementite in ferrite, that of the second consisted of partially spheroidised cementite in ferrite, and that of the third was pearlitic. In the untempered materials log δ, and therefore the damping, was smallest in the 1·1 per cent. carbon steel annealed at 680 deg. Cent. before cold drawing, while it was

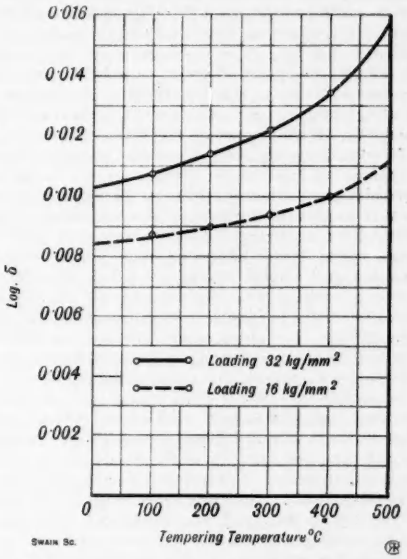


FIG. 8—*Effect of Tempering on Damping Capacity of Gongs made from Oil-hardened Wire.*

greatest in the 0·45 per cent. carbon specimen. The loads used in these determinations varied from 5 to 30 kg. sq. mm. Tempering at 300 deg. Cent. produced a large reduction in damping in the 1·1 per cent. carbon steel which had been annealed at 680 deg. Cent. These observations were therefore consistent with the results of the acoustic experiments on note duration.

The experiments reported above all referred to material which had been hardened by cold work, but in view of the well-known low damping capacity of steels hardened by quenching it was of interest to explore the acoustic properties of such materials. Parallel experiments were therefore made on gongs, the wires of which had been hardened by oil quenching from a temperature slightly above the A_3 transformation point and tempered at various temperatures. Fig. 7 shows the effect of tempering on the note duration, the steels being the same as those referred to in Fig. 6, and Fig. 8 the simultaneous changes in damping capacity in the 1·1 per cent. carbon steel.

In all except the 0·01 per cent. carbon material increase in tempering temperature produced a progressive increase in the length of note up to 500 deg. Cent.; above this point the duration began to diminish. The improvement in note duration on tempering was attributed to the relieving of hardening stresses, which requires higher temperatures than are necessary to remove internal stresses produced by cold work. The irregularities in the curves at 150, 250, and 300 deg. Cent. are considered to be due to transformations in the martensite and residual austenite. It seems doubtful, however, whether the accuracy of the determinations is sufficiently great for them to provide a reliable basis for this view. The higher carbon steels (0·63 and 1·1 per cent.) before tempering had a short note duration, superior only to that of the 0·01 per cent. carbon specimen, indicating the presence of high hardening stresses which accompanied their superior hardening capacity. Tempering, by releasing these stresses, yielded the best acoustic properties in the higher carbon steels.

Determinations of the damping capacity of the hardened 1·1 per cent. carbon steel, tempered at various temperatures, did not reveal the same correspondence between the logarithmic decrement and the note duration as was observed in the cold-drawn material; on the contrary, the damping, using loads of 16 and 32 kg. sq. mm., increased steadily with the tempering temperature. This apparent anomaly probably arose from the necessarily high alternating stresses used in determining the logarithmic decrement. Two opposing influences were at work in the gongs made from hardened steel; the increased hardness and tensile strength tended to reduce the damping, while the internal hardening stresses operated in the opposite direction. The latter were released by the alternating stresses, or were re-oriented in such a way that their influence on the damping was practically eliminated. The damping experiments therefore took account only of the hardness and tensile strength, and as tempering reduced both these it simultaneously caused an increase in the damping power. The different behaviour of work-hardened material is accounted for by the fact that internal stresses resulting from cold work are not re-oriented or removed by alternating stresses.

This example illustrates the value of an acoustic method for studying damping properties. It is clearly highly sensitive to internal stresses, and makes possible the observation of changes occurring under very small deformations. Since the method is essentially subjective in character, however, it is obviously necessary to check each observation as often as possible, so as to guard against errors arising from external influences, such as air movements, and from variations in aural sensitivity.

The Effect of Silver on the Softening Temperature of Copper.

THE softening temperature of cold-worked metals and alloys is dependent on many factors, such as method and amount of cold work, duration of the annealing operation, and grain size. It is also dependent on the composition, and it has, in fact, been shown both in the case of metals and alloys that the addition of comparatively small amounts of some elements very appreciably alters the temperature at which softening occurs. Quite a number of elements which include tin, silver, phosphorus, antimony, arsenic, and nickel, raise the softening temperature of copper. Of these, silver is one of the most useful for the purpose because of its comparatively neutral effect on other properties and its effect on the softening temperature is very marked. Incidentally, silver also has the effect of raising appreciably the limit of proportionality of copper, as observed recently by Hudson and McKeown. The raising of the softening temperature of a metal or an alloy is a matter of considerable practical interest and importance, especially when it can be effected by comparatively slight alloying additions without other properties of the material being adversely affected. In the case of copper the matter is of some consequence when the hard-worked metal is used in operations involving soldering or tinning, and in such applications copper carrying small amounts of silver is employed in preference to ordinary copper, because of its capacity to retain its strength and hardness at the temperatures to which it is exposed. Although Cæsar and Gerner, as long ago as 1916, showed that small amounts of silver raised the softening temperature of copper, there is still very little systematic information available on the matter. The subject has been recently investigated by H. C. Kenny and G. L. Craig, who have reported their results in Technical Publication No. 525 of the American Institute of Mining and Metallurgical Engineers. In the course of this work it has been shown that the softening temperature increases rapidly with a silver content up to 0·034 per cent. and then slowly with further increase in silver content; that cold-worked copper not containing silver softens completely in a few days at 150 deg. Cent., whereas copper containing 0·034 per cent. silver is not appreciably softened after a year at this temperature; and that as much as 0·14 per cent. silver has no adverse effect on the mechanical properties nor does it lower the electrical conductivity appreciably.

All the coppers tested were of the high conductivity tough-pitch variety with oxygen contents of the order of 0·05 to 0·07 per cent. For the softening tests with an annealing period of half-an-hour, coppers of varying silver contents, rolled with 20, 37, and 50 per cent. reduction in thickness were used. The softening temperature was taken as that at which the temperature curve became practically parallel to the hardness axis. There was no appreciable difference in the softening temperature of coppers rolled 37 and 50 per cent., but the softening temperature of copper rolled 20 per cent. was about 30 deg. Cent. higher than that of the more severely worked material. The softening temperature was raised rapidly at first by silver additions up to 0·034 per cent., when it was increased by approximately 130 deg. Cent., but further additions of silver increased it only slightly. Wires cold drawn from hot-rolled rods, $\frac{5}{16}$in. diameter to 0·020in. diameter, without annealing, were annealed for two-minute periods in a salt bath and tensile tests showed that the wires containing more than 0·052 per cent. silver were not completely softened even at 325 deg. Cent.

In the accompanying table are shown the results of experiments carried out to determine the effect of silver on the loss in strength of copper through tinning. For those experiments standard test pieces $\frac{1}{2}$in. wide and with a gauge length of 2in. were cut from coppers of varying silver content up to 0·13 per cent, cold rolled 20·6, 37·1, and 50·0 per cent. reduction in thickness. Tensile tests were made on samples before tinning and on samples which had been immersed for

ten seconds in a 60–40 lead-tin bath at 360 deg. Cent. The results obtained indicate that for material rolled 20·6 or 37·1 per cent. reduction in thickness, 0·037 per cent. silver is necessary to prevent softening on tinning, while for material rolled 50·0 per cent. reduction in thickness 0·052 per cent. silver is required.

Silver, per cent.	Before tinning.		After tinning.	
	Tensile strength, lb. per sq. in.	Elongation in 2in., per cent.	Tensile strength, lb. per sq. in.	Elongation in 2in., per cent.
		20·6 per cent.	reduction.	
0·018	40,900	13·9	39,700	14·0
0·022	39,300	13·5	38,800	15·8
0·037	38,900	16·4	39,000	16·5
0·052	40,100	16·9	40,000	16·6
0·070	40,300	15·3	39,800	13·8
0·087	40,800	12·5	40,200	14·5
0·100	39,800	16·0	40,000	13·5
0·119	40,400	14·6	40,600	14·4
0·130	40,900	12·3	41,000	14·8
		37·1 per cent.	reduction.	
0·018	49,200	2·5	44,600	7·0
0·022	48,400	2·2	46,100	5·4
0·037	49,400	2·75	47,700	3·8
0·052	49,600	2·9	48,100	4·5
0·070	50,700	2·75	48,700	4·3
0·087	50,500	2·75	49,200	3·3
0·100	50,800	2·5	49,400	3·3
0·119	51,100	2·4	49,200	4·4
0·130	51,300	2·6	49,000	3·4
		50·0 per cent.	reduction.	
0·018	55,000	1·6	34,200	38·4
0·022	54,600	1·3	34,200	36·5
0·037	54,600	1·6	45,700	7·6
0·052	54,500	1·6	51,200	3·2
0·070	55,500	1·6	52,600	3·0
0·087	55,500	1·6	52,700	2·3
0·100	55,500	1·6	53,000	2·1
0·119	56,000	1·6	53,300	2·4
0·130	56,000	1·6	53,700	2·3

It has been reported by Zeerleder and Bourgeois that cold-drawn silver-free copper gradually loses its strength at 80 deg. Cent. and when it is used for overhead transmission lines the heating effect of the current will cause softening of the wire. Kenny and Craig found that a sample of cold-drawn wire was completely softened in forty-one days at 80 deg. Cent. For determining the influence of silver on the rate of softening of copper at comparatively low temperatures over long periods of time they used wires cold-drawn, without any intermediate annealing, from $\frac{5}{16}$in. diameter to 0·050 in. diameter. Samples of these were maintained in a thermostatically controlled tube furnace at 150 deg. Cent. for as long as one year and at 200 deg. Cent. for 150 days, five wires of each composition being removed at intervals and their tensile strengths determined. At 150 deg. Cent. the wire containing 0·001 per cent. silver was completely softened in a few days and that containing 0·023 per cent. silver was appreciably softened towards the end of the test. With higher silver content the softening effect was very slight and the strength curves for wires containing more than 0·034 per cent. silver were apparently horizontal between 150 and 365 days. At 200 deg. Cent. none of the strength curves appeared to become horizontal. Even with as much as 0·13 per cent. silver some softening occurred. The rate of softening, however, clearly decreased with increasing silver content and the curves tended to become horizontal, an effect which might have been more evident in tests of more than 150 days' duration.

It will be noted from the tensile values given in the table that silver in amounts up to 0·13 per cent. has very little effect on the strength of high-conductivity tough-pitch copper and the electrical conductivity is only reduced from 101·0 per cent. of international standard to 100·1 by the presence of 0·132 per cent. silver.

Impact Testing of Cast Iron.

By J. G. PEARCE, M.Sc.

AN elaborate report has recently been issued under the above title by the American Society for Testing Materials.* The Sub-Committee concerned endeavoured to determine the utility of the several forms of impact test applied to cast iron, and with this object in view tested twenty-five sets of cast iron bars poured in greensand, with two exceptions, which were cast in dry sand. The bars were 1·4in. in diameter and 21in. long, irrespective of the suitability of the iron for this particular section. They were made in various foundries from metal used in current production and melted in various types of furnaces—cupola, rotary furnace, air furnace, and electric furnace. It is emphasised in the report that absolute values of the results obtained are of less importance than the relations between the static test results and the impact figures. There are tables giving casting data and chemical analyses, with static and impact tests. Complete transverse tests were taken on the bars, machined to 1·2in. diameter, at 18in. span (a standard B.S.I. test), to 0·5in. diameter at 5in. span, and to 8 mm. by 10 mm. section, 30 mm. span (a standard French test). Tensile tests were taken on bars machined to 0·5in. diameter; compression tests on 1·2in. bar; as well as shear, endurance limit, and hardness tests. No less than thirteen sets of impact tests were taken, as follows:—Charpy on 1·2in. diameter at 18in. span; Russell on 1·2in. diameter on both 18in. and 8in. span; Izod and Charpy tests on small specimens, both notched and unnotched; repeated blow Stanton test; a tensile impact; and four drop tests on 1·2in. bar, varying in weight of hammer, span, height of initial fall, and of subsequent increments. In general the impact tests of the single blow type are more common in laboratories and the drop test is more common in works, the conditions of the latter test being arranged to suit the material tested to give a result reasonably quickly.

The tables give, in addition, various calculations for physical constants, including the resilience, modulus of rupture, and modulus of elasticity from the transverse test results, the deflection D in the formula for the last mentioned being the total deflection at fracture. The report states that "the Committee is fully aware of the objections to this procedure, but uses these values (*i.e.*, rupture modulus and elastic modulus) as convenient indexes to stiffness and strength respectively."

With the mass of data resulting from the tests a considerable amount of ingenuity was expended in endeavouring to ascertain correlations between the impact strength and the static tests, but with very little success. Generally speaking, however, as would be expected, machines of different types employing similar operating principles, but differing in structural

* Report of Sub-Committee XV. of Committee A3 on Cast Iron, "Proceedings," A.S.T.M., Vol. 33, Part 1, 1933.

detail or in conditions of the test, gave results more or less concordant.

The following conclusions were reached :—

1. The very small specimens, such as are used for steel, are not reliable indices of the impact strength of cast iron. (Larger specimens were more satisfactory than smaller specimens.)
2. The Krupp-Stanton test gave relatively enormous differences in results from irons closely approaching each other in all other properties. It is difficult to see how this test means anything when applied to cast iron.
3. All notched specimens were unsatisfactory.
4. The tensile impact test does not seem to be a reliable index of impact value.
5. When using specimens 1·20in. in diameter the large pendulum machines of the Russell and Charpy type give results which follow closely the resilience measured in the transverse test.
6. The drop tests seem to indicate the impact resistance left in the bar after previous loading to a point just below the breaking point. A drop test, therefore, considered in the light of the transverse resilience, would appear to be a useful index of the relative amounts of plastic and elastic deflection.
7. The bending curve of the transverse test, if the test is accurately carried out, is capable of giving most of the information needed about cast iron, and the Brinell hardness determined over the whole cross section is the most valuable supplementary test.
8. The drop test machines differ greatly in their construction. They would probably become quite generally used if the machines were carefully standardised in all dimensions. The tests would then give valuable corroboration to the results of the transverse tests.
9. The Committee does not recommend at this time any move toward an impact requirement in specifications.

The conclusions 1 and 3 broadly confirm some earlier experience of the British Cast Iron Research Association, which showed that single blow impact tests on cast iron are not really satisfactory, whether notched or unnotched. This applies particularly to small specimens, but is still true even of larger bars than those commonly employed for impact tests on the steels. If graphite cavities in cast iron be regarded as notches, it will be clear that a sharp notch is not really necessary and for these reasons the Association adopted as an empirical impact test the Stanton bar with a rounded notch, on a repeated blow basis. This avoids the difficulties of the single blow test at the one extreme and the time involved in fatigue type tests at the other extreme. At the same time the repeated blow machine with a heavy hammer and large drop gives results comparable with the single blow machine and with a light hammer and a small drop gives results comparable with fatigue test.

At the invitation of the Sub-Committee, the writer subjected each set of bars to the repeated blow impact test, using a Stanton type test piece with a rounded notch and an energy of blow of 0·48 ft.-lb. There is an approximate relationship between the energy of blow (which can be varied within a limited range) and the number of blows to fracture. The blow adopted, therefore, tends to be on the knee of the hyperbolic-like curve connecting energy input and number of blows to fracture, and hence variations between different bars of the same material are reflected by rather large variations from the mean in the test results. The present writer, in a written communication on the Report, makes it clear that the test is regarded as wholly empirical, and that no attempt is made to relate it to other tests, or to use the results in design, or to suggest its use for specification or acceptance purposes. It is used merely as a laboratory method of assessing relative toughness or resistance to shock, and in this respect is quite satisfactory.

The present writer also points out that the wildness of the results on this test referred to by the Committee are in all probability actually due to variations within the material itself, and that a test which obscures these variations may defeat the object in view. Furthermore, it is necessary to discriminate between the intrinsic impact properties of the material and weakness due to unsoundness or internal stress, which may arise from the founding operation.

Mr. J. E. Hurst, in another communication printed in the Report, says that the resilience is the most satisfactory criterion of toughness in cast iron. This view is reflected in responsible quarters in this country, and is supported by conclusion 7. The present tendency in this country is to leave the impact test as a purely empirical test and to seek to develop a method of assessing toughness from a static test.

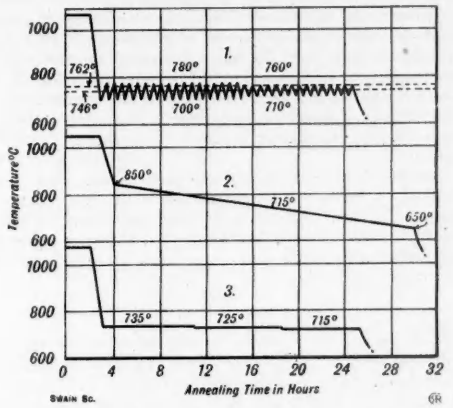
While the completion of such a formidable mass of data must be regarded with admiration, the results can only be described as disappointing. It is in the first place questionable whether an impact test which correlates with other static tests should be looked for. The whole virtue of the impact test lies in revealing conditions, as in the case of the steels, which the static tests do not reveal. Secondly, experience in this country has shown that, in a material where so many complex variables enter, in composition, in structure, and in methods of melting, comparatively little progress can be made by mass attack on the lines followed in this Report, and that it is preferable in the first place to tackle the problem on a laboratory basis, where the variables can be simplified, and where effects to be observed are, as far as possible, not masked by other changes.

Rapid Annealing of Blackheart Malleable Cast Iron.

ON account of the intermittent and protracted nature of the process, the annealing of blackheart malleable castings has always presented peculiar difficulties to the practical operator. It is not surprising therefore that a good deal of investigation, particularly in America, has been directed towards accelerating the process and at the same time assuring the production of uniform and malleable material.

One method of accelerating the process is to control the composition of the iron so as to favour the graphitisation of carbide occurring during the annealing period. The addition of silicon or any other element bringing about rapid graphitisation is, however, strictly limited, since, in general, excess renders sand castings grey and quite unsuitable for the production of good quality malleable. Melting, casting, and cooling conditions have been shown to have appreciable influence upon the rate of graphitisation during subsequent annealing, whilst the nature and pressure of the surrounding gas has also been stated to have some effect on the reaction. Of course, the temperature and rate of change of temperature during the anneal have a most important bearing upon the time required to complete the graphitising reaction, and

in this connection it is interesting to note that recently the German investigators, Merz and Schuster,* after a comprehensive survey of published data, have reached the conclusion that modification of the temperature cycle of the annealing treatment is the only practicable means by which the malleabilising of ordinary unalloyed iron may be accelerated. These investigators have therefore considered the mechanism and rate of graphitisation at different temperatures, and have evolved an annealing procedure claimed to be considerably more effective than any of the " short cycle " treatments hitherto suggested.



ANNEALING OF MALLEABLE CAST IRON

The top diagram represents the annealing temperature cycle recommended by Merz and Schuster, whilst the two lower diagrams show cycles based on the work of Hayes and Diederichs† and on recent patents of the General Electric Company. In order to obtain comparative results at lower temperatures, the initial process is in all cases carried out at about 1080 deg. Cent. This temperature is much higher than that generally used in practice, and seems to be advocated by Merz and Schuster for theoretical reasons and without reference to practical objections likely to be met with. In this connection it should be noted that an initial temperature of 925 deg. Cent. was used by Hayes and Diederichs in their own work. The object of the high annealing temperature is twofold—first, because equilibrium is more rapidly attained, and, secondly, because the graphite rosettes formed are more numerous, and therefore smaller, than at lower temperatures. For an iron containing carbon 2·46 per cent., silicon 0·96 per cent., manganese 0·52 per cent., sulphur 0·047 per cent., and phosphorus 0·08 per cent., the initial treatment at 1080 deg. Cent. may be reduced to as short a period as two hours.

According to the thermal equilibrium diagram for such an alloy, approximately 1·2 per cent. carbon remains in solid solution in contact with graphite at 1080 deg. Cent. It is therefore only by lowering the temperature that it is possible to graphitise this carbon and thus obtain malleability. As the temperature is lowered, the rate of reaction is diminished, so that the time required to graphitise the remaining 1·2 per cent. carbon is very much greater than the two hours necessary to complete the initial stage of the reaction. Referring to the equilibrium diagram, the solubility of carbon in gamma iron falls steadily to about 0·6 per cent. at the iron-graphite eutectoid point usually placed at 762 deg. Cent. Immediately below this temperature nearly all the remaining 0·6 per cent. carbon may be graphitised, but if sufficient time is not allowed, some of the carbon will be retained in solution and will be precipitated as metastable carbide at the pearlite eutectoid point, about 746 deg. Cent. This pearlite carbide may be decomposed in time, but if the temperature should fall too rapidly, it will persist in the structure and the final product will be hard and insufficiently ductile.

To achieve complete graphitisation, Hayes and Diederichs recommend cooling from 850 deg. to 715 deg. Cent. in fourteen hours, it being claimed that such treatment provides opportunity for rapid separation of graphite in excess of the iron-graphite eutectoid composition. The temperature is then lowered from 715 deg to 650 deg. Cent. in a further twelve hours, this slow cooling serving to graphitise any retained pearlite carbide. Since complete graphitisation may be obtained at the iron-graphite eutectoid point at 762 deg. Cent., it may be argued that a steady anneal just below this temperature would be more effective than the slow cool from 715 deg. to 650 deg. Cent. In the General Electric Company's procedure, the iron is cooled from 735 deg. to 715 deg. Cent. in three steps over a period of twenty-two hours. According to this authority, the carbon in excess of the iron-graphite eutectoid may be separated below the eutectoid temperature just as readily as above, and it is therefore considered unnecessary to cool slowly from the initial temperature to 735 deg. Cent.

It is generally accepted that equilibrium may be approached more rapidly by alternately heating and cooling over a small range than by maintaining a steady mean temperature. Although therefore the idea of the alternating anneal (*pendelglühung*) is not novel, Merz and Schuster have now found that, after the initial treatment at 1080 deg. Cent., subsequent graphitisation may be accelerated by alternately heating and cooling the iron through the critical range. At first the upper and lower limits of the temperature cycle (780 deg. to 700 deg. Cent.) are controlled so as to include both the iron-graphite and the iron-iron carbide eutectoid points, and it is shown that such a range is most effective in the first twelve-hour stage of the alternating treatment. The investigators prove, however, that in a completely graphitised iron, re-solution of carbon occurs quite rapidly as soon as the temperature is raised above that of the iron-graphite eutectoid. In order, therefore, to obtain complete graphitisation, the final ten-hour stage of the alternating anneal should be conducted below the iron-graphite eutectoid temperature, the range of the cycle (760 deg. to 710 deg. Cent.) being restricted so as to include only the iron-iron carbide eutectoid transformation.

The results of the three procedures may be compared in the table below extracted from the data published by Merz and Schuster.

The same iron was used in all experiments, the composition being as given above. It will be observed that the alternating anneal, put forward by Merz and Schuster, enables the combined carbon of the product to be reduced to a value well below that obtainable with either of the other two procedures. This is reflected in the high extension obtained in

* *Die Giesserei*, 20 (1933), pages 145–151, 173–181.

† "Trans.," Amer. Soc. Steel Treating, 6 (1924), page 491.

conjunction with a slight advantage in tensile strength.

Whilst the effectiveness of the alternating anneal appears to be definitely proved, it still remains to be

Investigators	Merz and Schuster.	Hayes and Diederichs.	General Electric Co.
Diagram reference	(1)	(2)	(3)
Annealing time, hours	25	30	25
Tensile strength, kilos. per square millimetre	44·0*	41·5	40·2
Extension on 60 mm. (12 mm. diameter test piece)	10·5*	6·1	4·0
Total carbon, per cent.	2·02	2·17	2·09
Graphitic carbon, per cent.	1·97	1·76	1·79
Combined carbon, per cent.	0·05	0·41	0·30
Proportion of pearlite in structure, per cent.	7·6	62·1	45·4

* Mean.

seen to what extent the data may be applied to the more efficient commercial annealing of blackheart malleable. Quite obviously, the relatively rapid rise and fall of temperature could not be obtained with the massive charges and large annealing furnaces in general use to-day. It might be argued, therefore, that it is unfair to compare the efficiency of the alternating treatment with that of other treatments in which there is practically no limit to the size of charge, &c. One function of investigation, however, is to show what can be done regardless of such limitations, and the results obtained by Merz and Schuster are certainly not rendered valueless because existing plant and technique can scarcely be adapted to meet the new requirements. If, in fact, the short duration and quality of product of the alternating anneal are considered to be of sufficient importance, there seems little doubt that new types of furnace and equipment will readily be developed to satisfy the new demands in regard to rapid heating and cooling of the charge. Such changes in design may also permit a higher initial annealing temperature than is generally considered practicable at present.

Effects of Centrifugal Action During Solidification.*

THE centrifugal casting of steel and iron, particularly the latter, is now a procedure of commercial importance. L. Quincy's treatment of the subject is of interest, but in no sense adds to our knowledge of the subject. The process is covered by patents which might be scanned by those intimately interested in the subject. Speed of rotation and rate and manner of input, together with the nature of the mould, are the principal factors in the process, and these are not quantitatively handled in the paper.

The first section discusses the author's conception of the principles of centrifugal action and the difficulties in the way of producing perfect castings, especially for thicker sections. Segregation and shrinkage effects have to be considered.

According to L. Quincy, it is necessary in casting a thick body that solidification should proceed steadily from the outside inwards, and for heavy masses this may demand a casting period beyond practical limitations for the ladle. He gives 6 mm. as a desirable thickness under normal conditions for the liquid steel layer at each stage, and in this case centrifugal force is considered to be only of use in distributing the metal.

During solidification a peripheric layer of solidified metal and a liquid centre portion is to be found. Cavities tend to form at the boundaries of the last frozen surface, but if the speed is sufficiently high the cavities will be displaced to lie along the central axis. If the speed is not sufficiently high in relation to the viscosity, cavities will lie excentrically. This is illustrated with experiments using castor oil.

Reactions.—Violent and prolonged mixing can be modified by regulation of the speed of rotation, and additions may be made in the ladle requiring time for reaction. An increased speed of rotation after filling the mould will lead to more intimate contact and increased speed of reaction. Products of reaction—silicates, &c.—tend to be carried to the centre. An illustration is given of spherical or egg-shaped slag particles (largely alumina) taken from near the centre of a mass.

Gases taken from a central hole in one case are claimed to have contained 93 per cent. hydrogen—hole 15 cu. dm. Such gas, at 1500 deg. Cent., would exert great pressure. It is suggested that the squeezing out of impurities and gases by high speeds is an important factor.

STRUCTURE.

Cast Iron.—The partly solidified portion has a reduced density and this tends to mix with the unchanged original liquid metal lying in the position of the core. This tends to equalise carbon contents and at the same time pressure effects tend to interfere with the growth of graphite flakes. Shrinkage of the core material tends to express graphite. Data are given showing the distribution of carbon and graphite in a given case.

Sulphur.—Sulphides tend to be carried towards the centre, but the usual segregation effects supervene, giving a distribution typified in Table I.

TABLE I.

Positions.	Sulphur.	Phosphorus.	Zones.
	Per cent.	Per cent.	
1	0·174	0·477	Central core
2	0·032	0·338	
3	0·036	0·351	
4	0·030	0·416	
5	0·030	0·572	Intermediate
6	0·062	0·429	Outside
7	0·064	0·390	
8	0·074	0·403	
9	0·068	0·429	

The advantages of this process as regards cast iron are summarised as follows :—

1. Diminution in total carbon content.
2. Pulverisation of eutectic graphite.
3. Restriction of the graphite.
4. Fineness of austenitic grains.
5. Desulphurisation and suppression of regions of MnS.
6. Elimination of phosphide eutectic network.

CAST STEEL.

The solidification of steel does not consist, like that of cast iron, of two stages, and only the phase

* L. Quincy, *Revue de Metallurgie*, February, 1934, pages 68–89.

corresponding to the hypoeutectic in cast iron has to be considered.

Carbon.—Here two principal zones, the outside zone and the central core, are found. For the same sized mass the difference between the two zones will not be so marked in steel as in cast iron because the solidification of steel takes place more rapidly owing to its high melting point. However, if the mass is considerable there will be important differences in the average carbon content of the outside zone and the core, as will be seen from Table II. As in the case of cast iron, the centrifugal force produces finer grain in the outer zone, and one photo-micrograph reproduced in the paper shows a flattening of the grains, which is analogous to that of the rolled condition.

Sulphur and Phosphorus.—Carbon, sulphur, and phosphorus behave identically. Owing to the relatively rapid solidification there is no considerable desulphurisation nor modification of the phosphorus content.

Manganese and Silicon.—As will be seen from Table II., there is little change in the distribution of the silicon and manganese.

TABLE II.

	Positions.						
	1	2	3	4	5	6	Cast.
	%	%	%	%	%	%	%
Carbon ..	0·72	0·66	0·63	0·45	0·44	0·44	—
Sulphur ..	0·045	0·044	0·030	0·018	0·019	0·018	0·033
Phosphorus	0·048	0·047	0·041	0·026	0·026	0·028	0·038
Silicon ..	0·35	0·34	0·33	0·32	0·32	0·30	0·33
Manganese	1·05	1·03	1·08	1·04	1·03	1·02	1·02
	Central core.			Outside zone			

As regards L. Quincy's conclusions, it is claimed to be shown that it is very difficult, if not impossible, to obtain solid or hollow castings by the normal method of centrifugal casting. It is stated that this new process, which is based on the centrifugation of a liquid mass, will permit of the production of thick hollow bodies in which the central hole corresponds to the pipe of the metal and which is characterised by the production of an exterior zone of superior quality. The process is covered by patents.

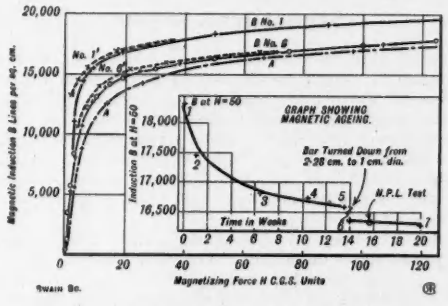
As will be seen from the above abstract, the article is of interest. The chief feature of L. Quincy's process would appear to be increased speed of rotation.

Magnetic Ageing.

IN a paper printed in the *Journal* of the Royal Technical College, Glasgow, Professor J. Muir, D.Sc., discusses an example of magnetic ageing shown by soft cast steel, in which the large induction B = 18300 for field H = 50 fell to 16500 in three months. The B—H curves were obtained from rods for which the ratio of length to diameter was only 30 to 35, so very large demagnetising corrections had to be applied to the magnetising fields produced by current in a long solenoid. These corrections were obtained from the results of Shuddemagen's experiments of 1907.

The first samples dealt with were two rods of steel about 26in. long and 1in. diameter, for which the percentage analysis, as given by the makers, was 0·12 C, 0·16 Si, 0·35 Mn, 0·046 P, 0·043 S. A single block of rectangular section had been cast on end, annealed by heating to 1000 deg. Cent., sawn up the middle, and turned into two rods by the steel makers. The magnetic properties were to be found by obtaining Rowland's reversal B-H curves, the magnetising force being taken high enough to ensure practical saturation of the steel. The main difficulty in the direct magnetic testing of rods is due to the large demagnetising effect of the magnetic poles induced in the specimen. In order to reduce this effect as far as possible, the rods were very carefully turned down in the College until the ratio of length to diameter was as large as possible, 30 or 35 to 1. Subsequently, the steel makers slightly altered the composition of the steel with the object of increasing the strength without injuring the magnetic quality. A percentage analysis of this new steel, obtained from a piece of the actual specimen tested, was 0·21 C, 0·42 Si, 0·45 Mn, 0·041 P, 0·042 S; the main change had been increase in silicon. The steel was supplied in the form of two rods which had been cast and annealed exactly in the same way as the first rods, and machined to the same size as before. On testing the first of these rods it was at once obvious from the kicks obtained from the ballistic galvanometer, that a decided improvement in magnetic quality had been effected. Curve B, No. 1, shows marked improvement over curve A.

The second rod of the new steel, tested a week later, did not show quite so great an improvement in magnetic quality. Points marked 1 and 2 on the inset ageing curve illustrate this difference sufficiently well without it being necessary to reproduce



MAGNETIC AGEING OF SOFT CAST STEEL

the complete B-H curve. The possibility of ageing was not thought of at the time, and the difference between the two rods was attributed to some casual flaw in the second rod, although both rods had been cut from the same casting.

The results were communicated to the makers, who sent a copy of the better graph to prospective steel users. About six weeks later a communication was received saying that the steel users had given this curve careful consideration, and that there must be some mistake in the testing, as the magnetic quality was too good for ordinary soft cast steel. The rod was therefore re-tested, with the disturbing result that the decided decrease in magnetic quality illustrated by point 3 of the inset diagram was obtained. All possible sources of error were carefully considered, and the reply given that there was no reason to doubt the accuracy of the tests made, but that the steel was showing a marked ageing with lapse of time. The prospective users were not satisfied

with this reply, and suggested that a test should be made at the National Physical Laboratory.

The N.P.L. required, for its test, a specimen exactly 1 cm. in diameter, so part of the rod in question was carefully turned down in the College from its initial diameter, 2·285 cm. Before doing this, two more magnetic tests were made, and points 4 and 5 on the inset graph show that ageing was still going on. Another test (No. 6), made immediately after turning down to 1 cm. diameter, showed a drop in induction (from 16600 to 16400 for field H=50), which might be accounted for by a hastening of the ageing process as a result of the machining; but there might be a difference in the mean magnetic quality of rods 1 cm. and 2·28 cm. in diameter. The complete B-H curve for this sixth test is shown in the diagram, and for comparison, the circles which lie almost exactly on that graph have been plotted to show the results of the N.P.L. test of the rod.

A seventh test was made after the rod had been returned by the N.P.L., and twenty weeks after the first test; ageing is still detectable, but has practically ceased. Although no cause can be given for the ageing effect observed, nor any suggestion made as to how the excellent initial magnetic properties of steel B could be made permanent, it has been thought that the publication of the observations made might serve some useful purpose.

The Welding of High-Tensile Structural Steels.

ALTHOUGH the welding of mild steel has now reached a stage at which results are in the highest degree efficient, and a perfect jointing of high-tensile steels can similarly be effected by fusion welding, there still remain problems affecting the characteristics of the weld which are peculiar to high-tensile steels. Probably the most important of these is the tendency towards air hardening exhibited to a greater or less degree by these steels.

Air hardening in this connection is a term that requires some definition. There is no difficulty in recognising a fully air-hardening steel which is martensitic in structure when cooled in air. Pearlite formation has been suppressed, and the carbon change completed at a lower temperature. At the other extreme is the low-carbon steel which is well known to be stronger in the normalised or air-cooled condition than after annealing and slow cooling. This strengthening is not air hardening in the same sense. The pearlite change has proceeded to completion, but the pearlite, which after slow cooling tends to be coarse or even granular, is more finely lamellar and the properties have usually been improved in every respect. The steel has a higher yield point and a higher ratio of yield point to maximum load. Its elongation may perhaps not be quite so good, but its reduction of area and impact figure are usually better. If, however, the composition is such that the whole of the carbon does not form pearlite during air cooling (the carbon change being partly suppressed and completed at a lower temperature with the formation of troostite or martensite), the steel is to some extent air hardening. Such an air-hardening tendency shows itself first as a fall in limit of proportionality, yield point, and yield ratio (probably the result of internal stress), and when this is distinctly shown, and before the elongation is appreciably affected, by a fall in impact figure also.

The composition of high-tensile steels which are to be used in the rolled condition is carefully chosen to avoid these detrimental consequences of air cooling from the rolling temperature. It must be remembered, however, that the cooling which follows welding is usually considerably quicker than air cooling, since heat is dissipated from the region of the weld, not only by radiation, but also by conduction to the adjacent colder metal. In a steel with air-hardening properties there is a risk that the boundaries of the weld may show a sharp peak in hardness with accompanying brittleness and high internal stress, which may be detrimental to the behaviour of the steel in fatigue. The tendency of the weld to assume these characteristics may, no doubt, to some extent be overcome by modifications in welding technique. Preheating of the steel in the vicinity of the weld reduces the temperature gradient and lessens the chilling effect, while the use of more generous fillets involving a larger mass of deposited metal provides a larger reserve of heat, the diffusion of which tempers the regions which have been rapidly cooled through the usual critical range. The hardness gradient near the weld is in this way flattened out and internal stresses relieved. Much can be done by these methods, but they introduce difficulties of their own—maintenance of uniformity in operation requires more skill, and there is an increased opportunity for oxidation to modify the composition of the weld metal itself.

A more favourable line of attack on the problem of welding high-tensile steel would appear to be for the metallurgist to produce high-tensile steels specially designed for use in welding construction. In steels for this purpose, elements such as manganese and chromium, which depend for their action mainly on their strengthening effect on the pearlite, would have to be restricted below their usual maximum limits in high-tensile steels generally, and so also would carbon; while more use could be made of silicon or of copper which exercise their strengthening effect on the ferrite. The carbide-forming elements, manganese, and chromium, tend to induce air hardening. Silicon has not the same effect, while copper, if used in suitable amounts, can, without any risk of inducing hardening on rapid cooling, compensate for a decrease in the amount of the usual hardening elements by the strengthening effect it exerts as a result of precipitation hardening in the ferrite.

In short, it seems desirable to avoid for welded structures compositions which approximate to the upper limits of carbon, manganese, and chromium permissible in high-tensile steel plate, and to use only compositions carefully restricted in these elements, but richer in copper or in silicon, both of which increase strength without inducing air hardening. Such control of composition is unnecessary in welded material which can be subsequently heat treated, and is, of course, impracticable in much repair work. In such instances as work on rails all the resources of welding technique are required to avoid undue hardening and embrittlement; but rather than introduce modifications of welding technique for the successful welding of high-tensile steel structures, it appears to be both practicable and desirable to control composition in such a way as to make it impossible for harmful air hardening to occur.